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Enhanced Machinability of Silicon Carbide via Microstructural Design

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The machinability of a heterogeneous silicon carbide with weak interphase boundaries, elongated grains, and high internal stresses is evaluated relative to a homogeneous control material with a well-bonded, equiaxed, and unstressed grain structure. Drilling and grinding rates for the silicon carbide are substantially enhanced by the microstructural heterogeneity—the weak boundaries enable easy grain-scale dislodgement in place of the more conventional macrofracture chipping mode of removal. At the same time, the residual machining damage in the machined surfaces is significantly less strength degrading in the heterogeneous material. Implications concerning the microstructural design of flaw-tolerant ceramics for enhanced machinability are considered.

I. Introduction

CERAMICS are increasingly being used in such engineering components as bearings, rotors, nozzles, valves, heat-exchangers, and substrates. Most of these components have complex shapes and require a good tolerance and surface finish. Such characteristics are usually achieved by machining, which is both costly and potentially damaging to the strength of the component.[1] In some cases (especially optical and electronic applications), machining damage may be so intrusive as to necessitate additional (and even more expensive) polishing operations. It is in this context that machining may be seen as a major limiting step in ceramics manufacturing.[2]

Accordingly, considerable effort has been devoted to increasing the efficiency of ceramic machining, primarily via improvements in cutting tool technology and process optimization.[3] In this communication we present an alternative, microstructural-design approach for enhancing the machinability of otherwise brittle ceramics. Our approach is similar to the one used to develop machinable glass-ceramics.[4] Here we focus on silicon carbide, a ceramic traditionally known for its innate hardness and brittleness. Specifically, we evaluate the enhanced machinability of a new form of heterogeneous silicon carbide fabricated using *in situ* processing, with a tailored microstructure consisting of coarse elongate grains and weak interphase boundaries.[5,6] A most dramatic manifestation of these microstructural features is a transition in damage mode in Hertzian contact loading: whereas the damage in a conventional homogeneous silicon carbide with well-bonded grain boundaries takes the familiar form of a well-defined cone-shaped macrocrack outside the contact circle, in the heterogeneous material it appears as a distributed zone of discrete microfaults below the contact circle.[6,7] In repeat loading, the damage zone in the heterogeneous material expands relatively quickly, leading ultimately to detachment of grain-scale material.[8,9] The same such fine-scale removal processes are observed in comparative tests on machinable glass-ceramics,[10] and account for the machinability of these materials.[4,11] The issue is whether the existence of a common microscale removal process in contact fatigue leads to enhanced machinability in the harder ceramic materials?

Here we report results of drilling and grinding tests for the heterogeneous silicon carbide relative to a homogeneous silicon carbide control. We confirm that the material removal rate of the heterogeneous material is enhanced, by a factor of about three, relative to its homogeneous counterpart. At the same time, we demonstrate that the severity of the machining damage is less intrusive, in particular in relation to strength. Implications concerning the design of ceramic microstructures for enhanced machinability are considered.

II. Experimental Procedure

Specimens of a heterogeneous silicon carbide were prepared to 97% density by *in situ* processing.[5] The microstructure consisted of elongate silicon carbide grains ≈ 3 μm thick and ≈ 25 μm long with 20 vol% yttrium aluminum garnet (YAG) crystalline second phase. The fracture mode in this material was intergranular, reflective of weak interphase boundaries with substantial thermal expansion mismatch stresses.[4,5]

Control silicon carbide specimens (again 97% density) were prepared from a relatively homogeneous commercial material (Hexoloy SA, Carborundum, Niagara Falls, NY). The grain structure was equiaxed,, with mean grain size ≈ 4 μm. In this material the fracture mode was transgranular, indicating well-bonded grain boundaries.[5,6]

Machining tests were conducted on the two silicon carbides, using rectangular coupons 6 mm × 6 mm × 3 mm cut from the batch materials. The prospective test surfaces were diamond polished to 1 μm finish. Low-speed grinding and drilling tests were then conducted as follows:

(i) *Grinding* tests were performed on an automatic lapping machine. Three silicon carbide specimens, two heterogeneous and one homogeneous, were glued symmetrically about the center of a circular metal plate, which was then attached to the head of the lapping machine via a universal joint. The specimens were ground on a 125-μm grit metal-bonded diamond wheel for 90 min (normal pressure 400 Pa and surface speed to 40 mm·s^{-1}, mineral oil lubricant). The thickness of material removed from each of the specimens was measured at 15-min intervals using a precision micrometer (accuracy ± 5 μm over grinding depth 300–1200 μm) and the removal rates (depth removed per unit time) thereby evaluated.

(ii) *Drilling* tests were conducted using a fixture described elsewhere.[12] A previously unused metal-bonded 120-grit diamond core drill bit (3.2-mm outer diam and 1.6-mm inner diam) attached to a milling machine was used to drill a blind hole in each specimen for 30 s under fixed operating conditions (normal load 34 N and drill rotation speed 30 Hz). Tap water was

pumped through the drill core for cooling. To accommodate bit wear,[12] the drilling sequence was alternated between the two materials, for five tests on each. The depth of each hole was measured by depth-of-focus adjustment on an optical microscope (accuracy ± 5 μm over hole depth 300–1200 μm), and the drilling rates computed.

A bonded-interface technique[13-15] was used to observe the subsurface grinding damage.[16] Two polished rectangular half-blocks were bonded across a common interface normal to the specimen test surface using a thin layer of adhesive. After grinding, the adhesive was dissolved in acetone and the half-blocks separated. The separated side surfaces were then gold-coated and viewed optically in Normarski illumination and scanning electron microscopy (SEM). Straightforward optical microscopy was used to examine the drilling damage.

Biaxial flexure tests were conducted to investigate the effect of grinding damage on strength. For these tests, specimens were prepared as disks of 20-mm diam and 2.5-mm thickness, with one surface diamond polished to 1 μm finish. The polished surfaces were then ground as described in (i) above, except that the grind time was limited to 9 min. To avoid environmental effects in the strength evaluation, a drop of silicone oil was placed on the ground surface, and the disks then broken in rapid loading (failure time <10 ms), with the ground surface on the tensile side. Strengths were calculated from the breaking loads.[17] Some disks were broken in their *un*ground (polished) states to establish a reference baseline.

III. Results

In the grinding tests on the silicon carbide specimens, the mean surface removal rate for the heterogeneous material was 0.20 μm·s^{-1} compared to 0.08 μm·s^{-1} for the homogeneous control, i.e., corresponding to an enhancement of almost a factor of 2.5. A similar degree of enhancement is evident in the drilling rate data plotted in Fig. 1, notwithstanding a sequential decrease in values as a result of drill-bit wear.[11]

Figure 2 shows section views of ground surfaces in the homogeneous and heterogeneous silicon carbides, using the bonded-interface technique. In both cases there is surface roughness ≈ 5 μm. The homogeneous material shows evidence of spasmodic chipping/spalling damage from traditional cone, median, and lateral cracking.[18-22] Remnant median cracks extending to a depth $c_f \approx 15$ μm are apparent. In the heterogeneous material the damage appears to occur by detachment of loosened grains. In this material there is no indication of any associated crack penetration into the underlayer.

The hole edges in the drilled surfaces showed signs of chipping damage. In the homogeneous material this damage was severe, indicative of the macrocrack chipping seen in Fig. 2(A)

but on a considerably larger scale. By contrast, the chipping damage in the heterogeneous material was relatively minor, similar to Fig. 2(B).

Strength data for silicon carbide disks are shown in Fig. 3, before and after grinding. The strength of the homogeneous material is substantially degraded, from 452 ± 52 MPa to 281 ± 13 MPa. The degraded strength corresponds to a flaw size $c_f = (T_0/\psi\sigma_F)^2 = 13$ μm, using baseline toughness $T_0 = 2.1$ MPa·m$^{1/2}$ (Ref. 6) and $\psi = 1.12\pi^{1/2}$ for linear surface flaws,[23] which compares with the measured value of 15 μm in Fig. 2(A). By contrast, the strength of the heterogeneous material hardly changes, i.e., 353 ± 45 MPa before, compared to 382 ± 12 MPa after. Thus, the heterogeneous material is much less susceptible to machining damage.

IV. Discussion

We have demonstrated that the machinability of silicon carbide can be markedly enhanced by incorporating heterogeneities, specifically weak interfaces, long grains and high thermal expansion mismatch stresses, into the microstructure. The micrographs in Fig. 2 indicate fundamentally different micromechanisms of material removal in our heterogeneous test material relative to a homogeneous control. In the homogeneous material, characterized by equiaxed grains and strong
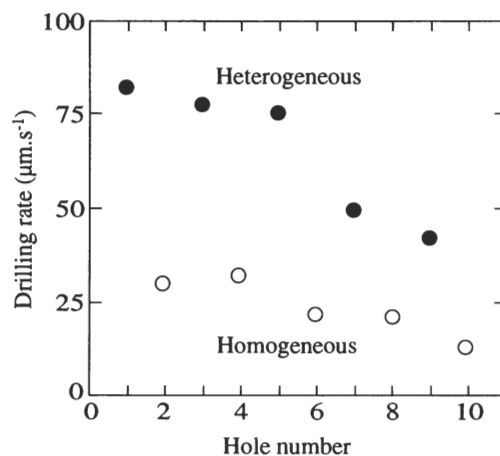
Fig. 1. Data showing drilling rates for homogeneous and heterogeneous silicon carbides. Holes are numbered in temporal sequence, alternating between the two materials. Note monotonic decrease in drilling rates with increasing number of holes, indicating progressive tool wear.
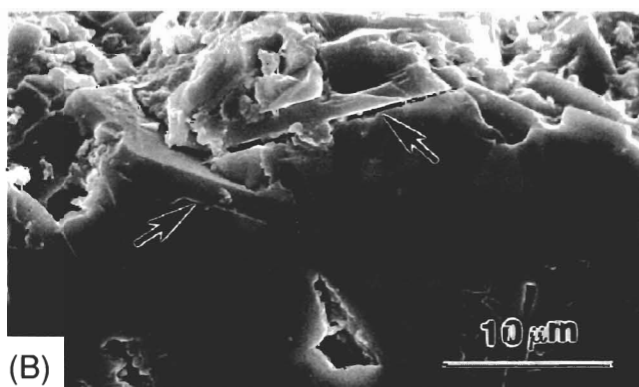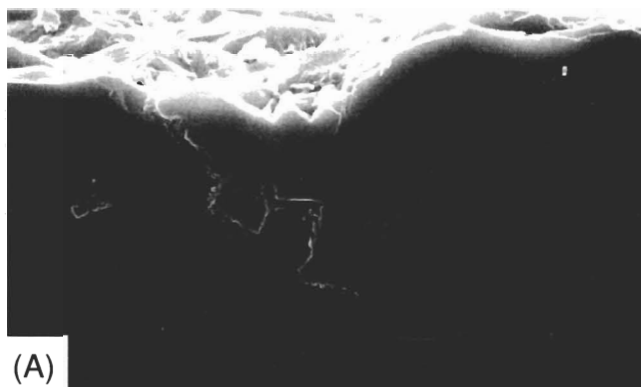
Fig. 2. Scanning electron micrographs, showing section views of grinding damage in (A) homogeneous and (B) heterogeneous silicon carbides. Arrows in (B) indicate local intergranular failures.
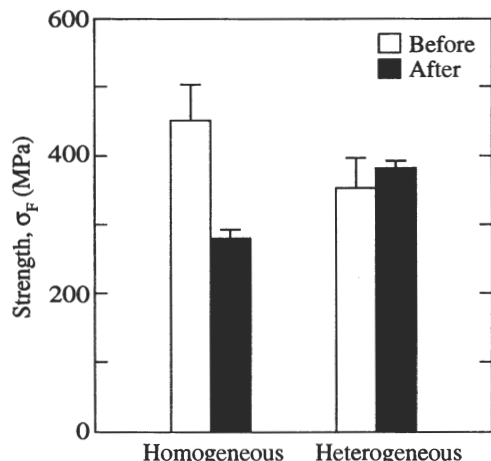
Fig. 3. Histogram showing strengths of homogeneous and heterogeneous silicon carbides, before and after surface grinding. Error bars are standard deviations, four specimens per condition.

intergranular bonding, the machining damage is consistent with a classical chipping mode, from mutual intersections of tensile cone or median/lateral transgranular macrocracks.[18,21] In our heterogeneous material, with its loosely bound platelet structure, any such incipient cracks are easily deflected from the maximum tensile stress trajectories by the weak interphase boundaries, effectively suppressing macrocrack formation.[22] Instead, facet-localized failures occur at the weak boundaries in the subsurface compression–shear zone, resulting in a distributed damage layer of discrete microfaults. Removal then occurs by detachment of individual grains from the contact surface, as observed in aluminas[25] and glass-ceramics.[26] Again, these inferred damage and removal modes are entirely consistent with those foreshadowed in our earlier cyclic Hertzian indentation studies on the same silicon carbide materials.[8,9]

Apart from an enhanced machining rate, the heterogeneous silicon carbide shows higher insensitivity to strength degradation from machining operations. As indicated above, the very weak-interface characteristic that promotes easy grain detachment suppresses macrocrack formation. Hence, whereas in Fig. 3 the homogeneous control material suffers substantial strength loss from grinding damage, the strength of the heterogeneous material is essentially unchanged.

It needs to be acknowledged that the degree of enhancement of machining properties is sensitive to specific machining conditions. The rates, and indeed micromechanisms, of surface removal in ceramics are notoriously variable, depending on such test parameters as tool hardness and geometry, test load, grinding or drilling speed, and wheel or bit wear. Accordingly, the relative heterogeneous/homogeneous performance values obtained for our silicon carbides can hardly be regarded as unconditional. Nevertheless, the present results do suggest that microstructural design can be an effective route to improved machinability of otherwise brittle ceramics, without compromising strength and toughness. Some of these issues will be explored elsewhere.[25–27]

References

[1]L. M. Sheppard, "The Challenges of Ceramic Machining Continue," *Am. Ceram. Soc. Bull.*, **71** [11] 1590–99 (1992).

[2]S. Jahanmir, "Machining of Advanced Materials," National Institute of Standards and Technology Special Publication 847. U.S. Government Printing Office, Washington, DC, 1993.

[3]S. Jahanmir, "Machining of Advanced Materials," National Institute of Standards and Technology Special Technical Publication 847. U.S. Government Printing Office, Washington, DC, 1993.

[4]K. Chyung, G. H. Beall, and D. G. Grossman, "Fluorophlogopite Mica Glass-Ceramics"; pp. 33–40 in *Proceedings of 10th International Glass Congress, No. 14.* Edited by M. Kunugi, M. Tashiro, and N. Saga. The Ceramic Society of Japan, Kyoto, Tokyo, Japan, 1974.

[5]N. P. Padture, "*In Situ*-Toughened Silicon Carbide," *J. Am. Ceram. Soc.*, **77** [2] 519–23 (1994).

[6]N. P. Padture and B. R. Lawn, "Toughness Properties of a Silicon Carbide with an *in Situ* Induced Heterogeneous Grain Structure," *J. Am. Ceram. Soc.*, in press.

[7]B. R. Lawn, N. P. Padture, H. Cai, and F. Guiberteau, "Making Ceramics 'Ductile'," *Science (Washington, DC)*, **263**, 1114–16 (1994).

[8]N. P. Padture and B. R. Lawn, "Fatigue in Ceramics with Interconnecting Weak Interfaces: A Study Using Cyclic Hertzian Contacts," *Acta Metall.*, in press.

[9]N. P. Padture and B. R. Lawn, "Contact Fatigue of a Silicon Carbide with a Heterogeneous Grain Structure," *J. Am. Ceram. Soc.*, in press.

[10]H. Cai, M. A. S. Kalceff, B. M. Hooks, B. R. Lawn, and K. Chyung, "Cyclic Fatigue of a Mica-Containing Glass-Ceramic at Hertzian Contacts," *J. Mater. Res.*, in press.

[11]C. K. Chyung, G. H. Beall, and D. G. Grossman, "Microstructures & Mechanical Properties of Mica Glass-Ceramics"; pp. 1167–94 in *Electron Microscopy and Structure of Materials.* Edited by G. Thomas, R. M. Fulrath, and R. M. Fisher. University of California Press, Berkeley, CA, 1972.

[12]S. Jahanmir and H. Liang, "Chemomechanical Effects of Boric Acid in Core-Drilling of Alumina"; pp. 101–13 in *Machining of Advanced Materials*, Vol. National Institute of Standards and Technology Special Publication No. 847. Edited by S. Jahanmir. U.S. Government Printing Office, Washington, DC, 1993.

[13]F. Guiberteau, N. P. Padture, and B. R. Lawn, "Effect of Grain Size on Hertzian Contact in Alumina," *J. Am. Ceram. Soc.*, **77** [7] 1825–31 (1994).

[14]H. Cai, M. A. Stevens Kalceff, and B. R. Lawn, "Deformation and Fracture of Mica-Containing Glass–Ceramics in Hertzian Contacts," *J. Mater. Res.*, **9** [3] 762–70 (1994).

[15]H. H. K. Xu, L. Wei, N. P. Padture, B. R. Lawn, and R. L. Yeckley, "Effect of Microstructural Coarsening on Hertzian Contact Damage in Silicon Nitride," *J. Mater. Sci.*, in press.

[16]H. H. K. Xu and S. Jahanmir, "Simple Technique for Observing Subsurface Damage in Machining of Ceramics," *J. Am. Ceram. Soc.*, **77** [5] 1388–90 (1994).

[17]D. B. Marshall, "An Improved Biaxial Flexure Test for Ceramics," *Am. Ceram. Soc. Bull.*, **59** [5] 551–53 (1980).

[18]B. R. Lawn and T. R. Wilshaw, "Indentation Fracture: Principles and Applications," *J. Mater. Sci.*, **10** [6] 1049–81 (1975).

[19]A. G. Evans and T. R. Wilshaw, "Quasi-Plastic Solid Particle Damage in Brittle Materials," *Acta Metall.*, **24**, 939–56 (1976).

[20]M. V. Swain, "Microfracture about Scratches in Brittle Solids," *Proc. R. Soc. Lond., A*, **366**, 575–97 (1979).

[21]D. B. Marshall, A. G. Evans, B. T. K. Yakub, J. W. Tien, and G. S. Kino, "The Nature of Machining Damage in Brittle Materials," *Proc. R. Soc. Lond., A*, **385**, 461–75 (1983).

[22]O. O. Ajayi and K. C. Ludema, "Surface Damage of Structural Ceramics: Implications for Wear Modelling," *Wear*, **124**, 237–57 (1988).

[23]B. R. Lawn, *Fracture of Brittle Solids.* Cambridge University Press, Cambridge, U.K., 1993.

[24]F. Guiberteau, N. P. Padture, H. Cai, and B. R. Lawn, "Indentation Fatigue: A Simple Cyclic Hertzian Test for Measuring Damage Accumulation in Polycrystalline Ceramics," *Philos. Mag. A*, **68** [5] 1003–16 (1993).

[25]H. K. Xu and S. Jahanmir, "Microfracture and Material Removal in Scratching of Alumina," *J. Mater. Sci.*, in press.

[26]H. H. K. Xu and S. Jahanmir, "Scratching and Grinding of a Machinable Glass-Ceramic with Weak Interfaces and Rising *T*-Curve," *J. Am. Ceram. Soc.*, in press.

[27]H. H. K. Xu, N. P. Padture, and S. Jahanmir, "Effect of Microstructure on Material-Removal Mechanisms and Damage Tolerance in Abrasive Machining of Silicon Carbide," *J. Am. Ceram. Soc.*, in press. □